

Creep at Low Stresses: An Evaluation of Diffusion Creep and Harper–Dorn Creep as Viable Creep Mechanisms

TERENCE G. LANGDON

High-temperature creep experiments often reveal a transition at very low stresses to a region where the stress exponent is reduced to a value lying typically in the range of ~ 1 to 2. This region is generally associated with the occurrence of a new creep mechanism, such as grain-boundary sliding, diffusion creep, and/or Harper–Dorn creep. Several recent reports have suggested that diffusion creep and Harper–Dorn creep may not be viable creep mechanisms. This article examines these two processes and demonstrates that there is good evidence supporting the occurrence of both creep mechanisms under at least some experimental conditions.

I. HISTORICAL BACKGROUND

CREEP deformation refers to the unrecoverable plastic strain occurring in a material when it is subjected to a constant applied stress (or a constant applied load) over an extended period of time. Creep processes are diffusion-controlled, and they become of particular importance in materials experiencing extensive periods of time at elevated temperatures, where these high temperatures are generally above $\sim 0.4 T_m$, where T_m is the absolute melting temperature of the material.

The small-scale industries of the 19th century tended to operate at relatively low temperatures so that the occurrence of any creep deformation in mechanical parts was generally neither appreciated nor of significant magnitude to seriously impair the industrial operation. However, this situation began to change in the very early days of the 20th century when there was a concerted effort to increase the operating temperatures, and therefore the overall efficiency, of conventional working plants, such as steam boilers. It is not surprising, therefore, that the first scientific publication dealing exclusively with creep deformation should appear almost exactly 100 years ago in the classic report by Phillips^[1] on the creep deformation occurring, as a function of time, in materials as diverse as Indiarubber, glass, and metal wires. This early article was followed initially by other limited publications on creep, most notably the early report by Andrade^[2,3] claiming a $t^{1/3}$ law in which the creep strain increases with time, t , raised to the third power, until ultimately, within the last 30 years, there appear annually a plethora of reports describing creep deformation in a very wide range of metallic and nonmetallic materials.

An important change in direction appeared in the 1950s with the classic work of Dorn and his colleagues, where a phenomenological approach was developed and careful

experiments were undertaken to determine the precise functional relationship between the steady-state creep rate and external experimental parameters such as stress and temperature. This latter approach had two very significant advantages over the earlier attempts to develop constitutive relationships as in the classic Andrade $t^{1/3}$ law. First, the approach, when combined with theory, permitted an assessment of the precise atomistic processes occurring during creep deformation, and thus it led to the concept of specific and well-defined rate-controlling creep mechanisms. Second, the approach provided, for the first time and over at least a reasonable range of experimental conditions, a predictive capability of the effect of changes in the operating stresses and temperatures. The many publications of this era are well documented in the creep literature, and they culminated in an extended and comprehensive overview of the creep behavior of a wide range of metals and metallic alloys.^[4]

This article follows on from this more recent approach, and it is concerned specifically with the significance of steady-state creep and the interpretation of rate-controlling creep mechanisms at very low stress levels. As will be demonstrated, although the atomistic mechanisms of creep are now well-documented at intermediate and high stresses, the difficulties of accurately recording extremely slow rates of deformation have necessarily led to uncertainties and ambiguities in precisely interpreting the creep behavior at these very low rates.

II. REGIONS OF CREEP BEHAVIOR

An important characteristic of high-temperature creep is that the steady-state creep rate, $\dot{\epsilon}$, generally varies with the applied stress, σ , the absolute temperature, T , and the grain size of the material, d , through a relationship of the form

$$\dot{\epsilon} = \frac{ADG\mathbf{b}}{kT} \left(\frac{\mathbf{b}}{d}\right)^p \left(\frac{\sigma}{G}\right)^n \quad [1]$$

where D is the appropriate diffusion coefficient ($= D_0 \exp(-Q/RT)$), where D_0 is a frequency factor, Q is the activation energy, and R is the gas constant), G is the shear modulus, \mathbf{b} is the Burgers vector, k is Boltzmann's constant, p and n are the inverse grain size exponent and the stress exponent, respectively, and A is a dimensionless constant. It follows from Eq. [1] that a plot on logarithmic axes of $\dot{\epsilon}$ against σ

TERENCE G. LANGDON, Professor, is with the Departments of Aerospace & Mechanical Engineering and Materials Science, University of Southern California, Los Angeles, CA 90089-1453.

This article is based on a presentation made in the workshop entitled "Mechanisms of Elevated Temperature Plasticity and Fracture," which was held June 27–29, 2001, in San Diego, CA, concurrent with the 2001 Joint Applied Mechanics and Materials Summer Conference. The workshop was sponsored by Basic Energy Sciences of the United States Department of Energy.

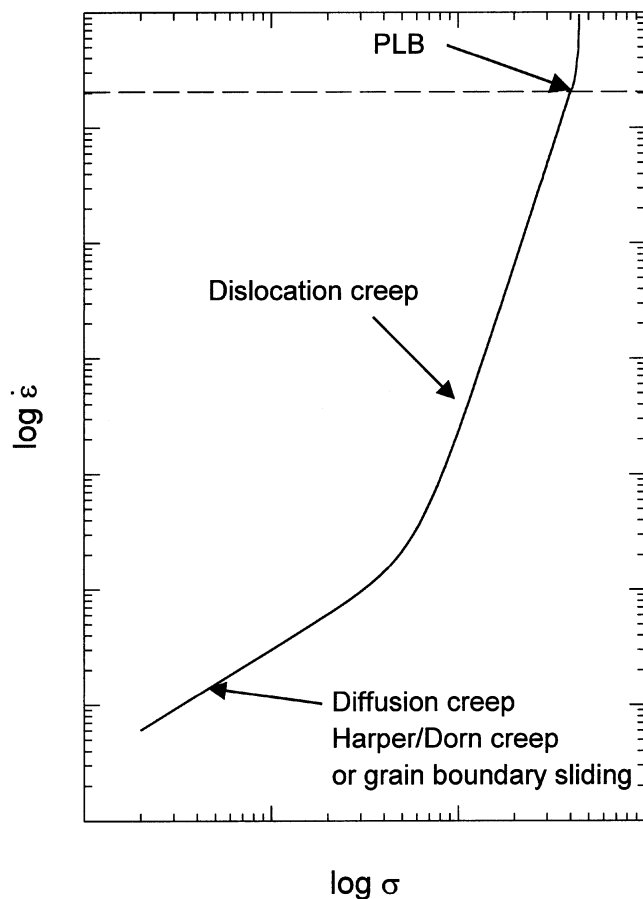


Fig. 1—Schematic illustration of the variation in a logarithmic plot of the steady-state creep rate, $\dot{\epsilon}$, with the applied stress, σ , for a typical polycrystalline metal: the plot reveals an extensive region where dislocation creep is dominant at intermediate stresses, PLB at very high stresses, and a transition to a region having a lower slope at low stresses due to the occurrence of diffusion creep, Harper–Dorn creep, or grain-boundary sliding.

will lead to the experimental datum points falling along a line having a slope equal to the value of n .

Figure 1 shows a schematic illustration of representative creep behavior in a typical metal. In general, the creep behavior, when analyzed in terms of the steady-state creep rates, may be divided into three separate regions.

First, there is a wide stress range at intermediate stresses where the datum points usually fall along a line having a slope within the range of ~ 3 to 6 . This region is associated unambiguously with some form of dislocation creep in which the creep strain is accrued from the intragranular movement of dislocations through the processes of glide and climb. Generally, dislocation climb is rate-controlling in pure metals, and n is in the range of ~ 4.5 to 6 with the precise value dependent upon the stacking-fault energy of the material.^[4] In metallic alloys, the rate-controlling process may be dislocation glide with $n = 3$ in association with the dragging of solute atom atmospheres. The occurrence of dislocation glide as a rate-controlling process in alloys may lead to deviations from a straight line at these intermediate stresses due to transitions to climb-controlled behavior either at the lower stresses where climb is slower than glide^[5] and/or at the higher stresses where the dislocations break away from the solute atom atmospheres.^[6]

Second, there is a deviation from linearity at very high stresses due to the advent of power-law breakdown (PLB). This deviation is well documented in the creep literature, and it represents, in effect, a transition from the diffusion-controlled regime associated with high temperatures to a thermally activated regime analogous to low temperature flow.^[7] It has been shown that PLB occurs in fcc metals at a normalized strain rate given by^[8]

$$\frac{\dot{\epsilon}}{D} \approx 10^{13} \text{ m}^{-2} \quad [2]$$

or alternatively at a normalized stress of the order of^[9]

$$\frac{\sigma}{G} \approx 5.5 \times 10^{-4} \quad [3]$$

Third, if the experiments are extended to cover very low stress levels, there is generally a transition to a new region at the lowest stresses where the value of n decreases and lies typically within the range of ~ 1 to 2 . Although the precise origin of this change of slope is usually not well understood, the region of low n is generally attributed to the advent of a new creep mechanism, such as diffusion creep, Harper–Dorn creep, and/or grain-boundary sliding. In the following sections, the uncertainties and ambiguities in this low stress region are examined in detail.

III. CREEP PROCESSES AT VERY LOW STRESSES

Three separate creep mechanisms may occur at very low stresses.

A. Diffusion Creep

Under the action of an external stress, there is a depletion of vacancies along those grain boundaries experiencing a compressive stress and a corresponding excess of vacancies along those grain boundaries experiencing a tensile stress. Diffusion creep refers to the stress-directed flow of vacancies that takes place in order to restore an equilibrium condition. This process is illustrated schematically in Figure 2(a), where it is apparent that vacancy flow associated with the idealized square grain may occur either through the crystalline matrix in the process known as Nabarro–Herring diffusion creep^[10,11] or along the grain boundaries in the process known as Coble diffusion creep.^[12] It can be shown theoretically that there is a value of $n = 1$ in Eq. [1] for both of these processes but $p = 2$ and $D = D_\ell$ for Nabarro–Herring creep whereas $p = 3$ and $D = D_{gb}$ for Coble creep, where D_ℓ and D_{gb} are the diffusion coefficients for lattice self-diffusion and grain-boundary diffusion, respectively. Figure 2(b) illustrates the effect of diffusion creep in a material containing particles: as a consequence of the elongation of the individual grains, there is a buildup of particles along those grain boundaries lying more nearly parallel to the tensile axis whereas denuded zones, having a depletion of particles, form preferentially on those grain boundaries lying more nearly perpendicular to the tensile axis.

B. Harper–Dorn Creep

Harper–Dorn creep was first reported by Harper and Dorn^[13] in a series of classic experiments in which tensile

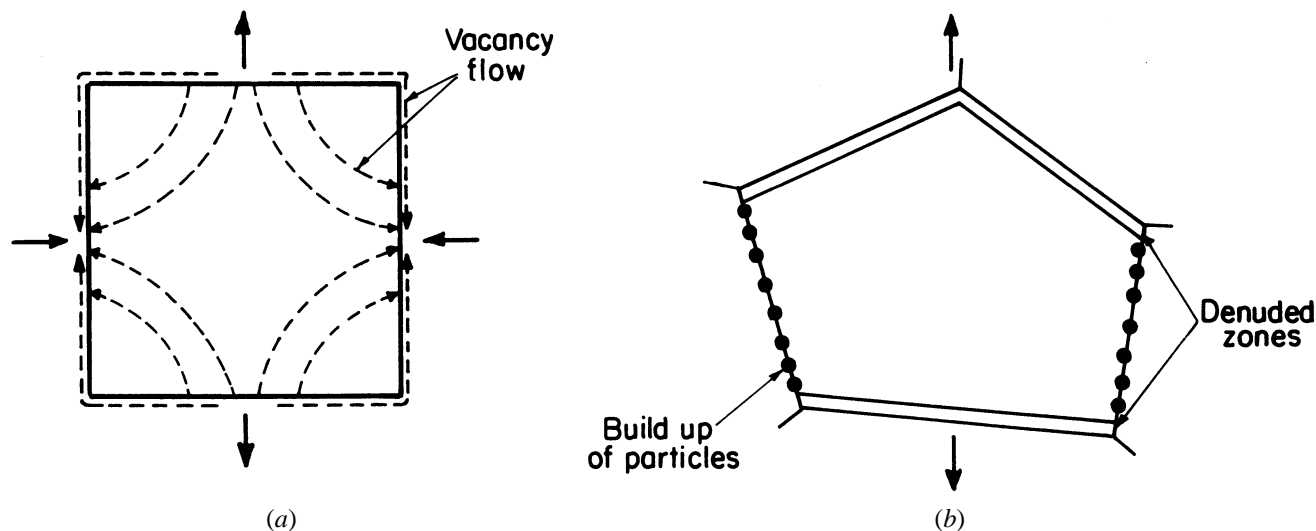


Fig. 2—The principle of diffusion creep: (a) showing vacancy flow through the grains (Nabarro–Herring creep) or along the grain boundaries (Coble creep) and (b) showing the consequent buildup of particles on longitudinal grain boundaries and the formation of denuded zones due to the depletion of particles along transverse grain boundaries.

creep tests were conducted on single crystals and polycrystalline samples of pure aluminum at temperatures very close to the melting temperature. The results showed a creep region at very low stresses with $n = 1$ and $D \approx D_\ell$ but with at least two important characteristics that appeared inconsistent with conventional Nabarro–Herring creep. First, the measured experimental creep rates were significantly faster, by more than two orders of magnitude, than those predicted theoretically by Nabarro–Herring creep. Second, identical rapid creep rates were recorded for both polycrystalline samples with a grain size of $d \approx 3$ mm and for single crystals, thereby implying that $p \approx 0$ in Eq. [1]. There have been numerous subsequent reports appearing to confirm the occurrence of Harper–Dorn creep in a number of materials and much of this information was summarized in a review.^[14]

C. Grain-Boundary Sliding

There is no doubt that the individual grains of a polycrystalline matrix may become displaced with respect to each other during high temperature creep. These displacements are easily recorded using marker lines, and the surface offsets associated with grain-boundary sliding have been measured and analyzed in many different experiments.^[15] An important characteristic of grain-boundary sliding is that the grains retain essentially their original shape so that, in the extreme of very high strains, there is an increase in the total number of grains lying along the tensile axis. The high tensile ductilities of superplastic flow are a classic example of the occurrence of grain-boundary sliding without any significant changes in the shapes of the individual grains.^[16] Following the conventional notation,^[17] this process is henceforth designated Rachinger grain-boundary sliding.^[18]

A simple consideration of Rachinger sliding shows that it cannot occur in a polycrystalline material without the additional occurrence of some concomitant accommodation within the grains. In practice, Rachinger sliding is accommodated by slip in the grains, and the experimental evidence suggests that the rate of sliding along a grain boundary is controlled, ultimately, by the rate at which dislocations are

able to move into the adjacent grain in the accommodation process. This approach leads in a simple way to a unified model for Rachinger sliding in which the precise form of Eq. [1] is dependent upon the nature of the accommodation. Thus, in large-grained polycrystalline metals, the accommodating dislocations move into the adjacent grains and impinge upon the subboundaries formed during creep and the rate of sliding is controlled by the rate of climb of the leading dislocations into these subgrain boundaries; whereas in materials with very small grain sizes, as in superplasticity, no subgrains are formed, and the rate of sliding is controlled by the rate of climb of the leading dislocations into the opposite grain boundary. These two situations are illustrated schematically in Figures 3(a) and (b), respectively.^[19] In Figure 3(a), sliding occurs through dislocation movement on the boundary of a very large grain, the sliding is blocked by a ledge at A, and the accommodating dislocations move into the adjacent grain and pile up at the first subgrain boundary; whereas in Figure 3(b), there are no subgrains within the very small grains in superplasticity so that sliding between grains B and C leads to a stress concentration at the triple junction D that is accommodated by dislocations moving across the grain and climbing into the opposite grain boundary. The critical grain size associated with the transition between these two forms of Rachinger sliding is when the grain size, d , is equal to the equilibrium subgrain size, λ . Thus, Figure 3(a) applies at large grain sizes when $d > \lambda$, whereas Figure 3(b) applies at very small grain sizes when $d < \lambda$, where it is well established that the value of λ under equilibrium creep conditions is given by^[4]

$$\frac{\lambda}{b} = \zeta \left(\frac{\sigma}{G} \right)^{-1} \quad [4]$$

where ζ is a constant having a value close to ~ 20 .

Using this approach, it can be shown theoretically that Rachinger sliding can be represented by Eq. [1] with $n = 3$, $p = 1$, and $D = D_\ell$ for large grain sizes in high temperature creep when $d > \lambda$, whereas for small grain sizes in superplasticity, when $d < \lambda$, the values are $n = 2$, $p = 2$, and $D =$

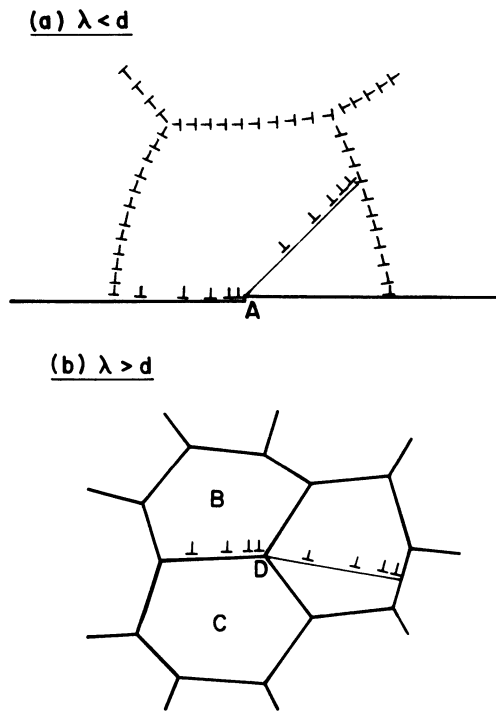


Fig. 3—The principle of grain-boundary sliding: (a) as in conventional creep when the grain size, d , is larger than the equilibrium subgrain size, λ , and the accommodating intragranular slip impinges on the subgrain boundaries within the grains; and (b) as in superplasticity when the grain size, d , is smaller than the equilibrium subgrain size, λ , and the accommodating intragranular slip impinges on the opposite grain boundaries.

D_{gb} .^[19] Careful analyses of extensive mechanical data for two typical superplastic alloys, the Zn-22 pct Al eutectoid and the Pb-62 pct Sn eutectic, have suggested that superplasticity occurs only under conditions where $d < \lambda$.^[20] This is illustrated by the experimental deformation mechanism map for the Zn-22 pct Al alloy shown in Figure 4 for a testing temperature of 503 K, where region II is the superplastic region, region III occurs at higher stresses and represents the transition to a nonsuperplastic behavior, region I is a low stress region now known to be associated with the presence of impurities in the boundaries,^[21,22] and the regions of Nabarro-Herring and Coble diffusion creep were inserted based on the theoretical relationships for these two processes. This map depicts the experimental data in the form of the normalized grain size, d/b , vs the normalized shear stress, τ/G , where τ is the shear stress, and a broken line is superimposed representing Eq. [4] with $d = \lambda$ and with the applied stress replaced by the shear stress. It is apparent that this line is essentially coincident with the experimental transition from nonsuperplastic flow at the higher stresses to superplastic flow at the lower stresses, thereby providing strong support for the occurrence of Rachinger sliding with $n = 2$ in superplasticity at grain sizes where no subgrains are able to form.

The occurrence of dislocation slip within the grains as an accommodating process in superplasticity has been demonstrated in two separate ways. First, through experiments showing that the strains within the individual grains of a Pb-62 pct Sn alloy are oscillatory in nature during superplastic flow, and they make no net contribution to the total

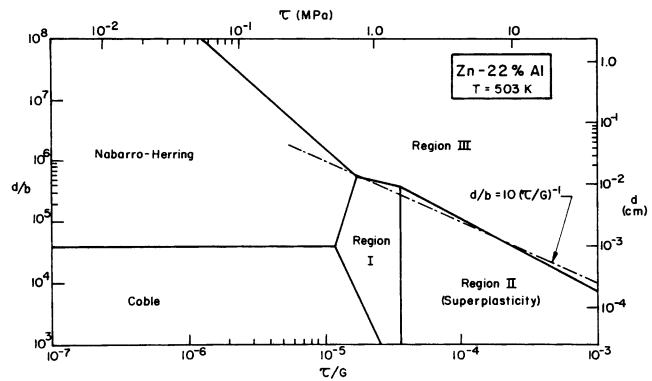


Fig. 4—Deformation mechanism map of normalized grain size vs normalized shear stress for Zn-22 pct Al at 503 K showing that superplasticity occurs when the grain size is smaller than the equilibrium subgrain size defined by Eq. [4].^[20]

elongation of the specimen.^[23] Second, by directly measuring the densities of intragranular dislocations trapped in coherent twin boundaries in a superplastic Cu alloy.^[24]

It is important to note that Rachinger sliding generally makes a relatively minor contribution to the overall strain of a polycrystalline material when the grain size is large as in conventional creep, but in superplasticity, when the grain size is very small and $d < \lambda$, the evidence suggests that Rachinger sliding accounts for essentially all of the deformation.^[16]

IV. A PROCEDURE FOR DISTINGUISHING BETWEEN DIFFUSION CREEP AND HARPER-DORN CREEP

The occurrence of Rachinger grain-boundary sliding is easily revealed in creep experiments through the offsets in surface marker lines that occur at the points where the lines impinge upon the grain boundaries. However, it is less easy to distinguish experimentally between the occurrence of diffusion creep and Harper-Dorn creep since Nabarro-Herring creep and Harper-Dorn creep both give $n = 1$ and they both have an activation energy for creep equal to the value for lattice self-diffusion, Q_ℓ . These similarities, combined with the experimental problems of measuring very slow creep rates over long periods of time, have led to many questions concerning the viability of these two creep mechanisms. Thus, reports have been published suggesting there is no good experimental evidence supporting the occurrence of diffusion creep^[25,26] or Harper-Dorn creep,^[27] and these reports were followed by equally vigorous publications defending the occurrence of both diffusion creep^[28] and Harper-Dorn creep.^[29] A detailed listing of these various claims and counterclaims is beyond the scope of this article but a tabulated summary of many of the reports is given in a recent review.^[30] These contradictory reports serve to emphasize the need to develop criteria that may be used to distinguish unambiguously between diffusion creep and Harper-Dorn creep while recognizing that both processes lead to an elongation of the individual grains, and therefore, at least from a macroscopic viewpoint, they are essentially equivalent.

In seeking procedures for distinguishing between these two mechanisms, it is important to note that intragranular

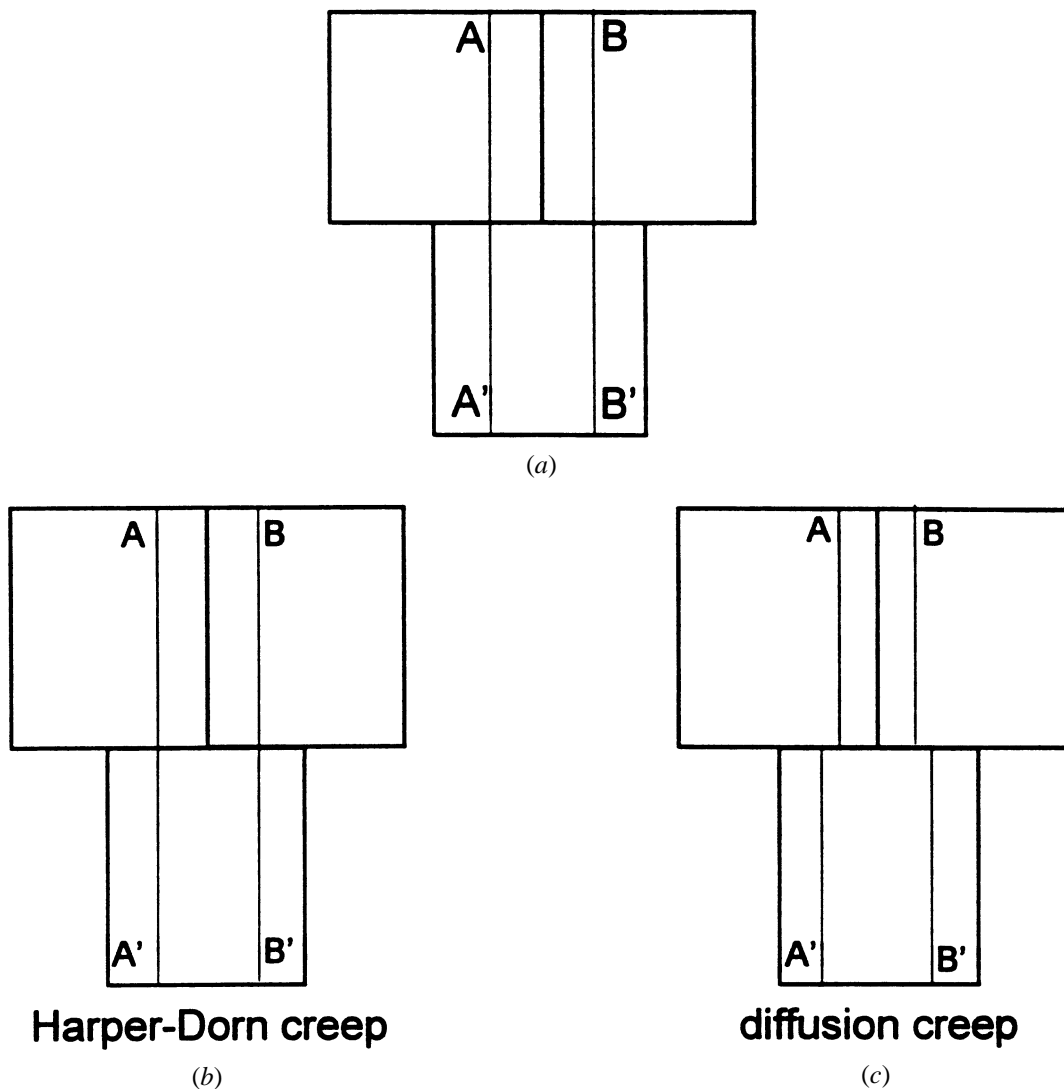


Fig. 5—Appearance of three grains in a polycrystalline matrix (a) before tensile creep with longitudinal marker lines AA' and BB', (b) after Harper–Dorn creep, and (c) after diffusion creep: the tensile axis is vertical.

dislocation processes are unique because they require no accommodation process, and thus they are capable of accounting fully, and exclusively, for all of the deformation occurring within a polycrystalline matrix. The dislocation creep processes occurring over a wide range of intermediate stresses fall into this category, as illustrated in Figure 1, and also the region of Harper–Dorn creep where, although the precise deformation mechanism is not known, it is generally attributed to some form of intragranular dislocation process. On the other hand, diffusion creep through vacancy flow leads to an elongation of the grains along the tensile axis through both a deposition of atoms on the transverse grain boundaries and a removal of atoms from the longitudinal grain boundaries. Thus, diffusion creep requires accommodation by concomitant grain-boundary sliding whereas Harper–Dorn creep needs no accommodation.^[31] This situation is illustrated schematically in Figures 5 and 6, where the tensile axis is vertical, and the samples contain marker lines either parallel to or perpendicular to the tensile axis, respectively.^[32] Figure 5(a) shows the situation before creep for three grains in a polycrystalline matrix containing two marker lines, AA' and BB', and Figures 5(b) and (c) show

the same grains after deforming by Harper–Dorn creep and diffusion creep, respectively. In both situations, the total strain is the same, and the grains are elongated by the same amount but in Harper–Dorn creep there is no accommodation process, and the marker lines remain continuous across the horizontal interfaces whereas in diffusion creep atoms are removed at the longitudinal boundary between the two upper grains and this leads to sharp offsets in the marker lines where they impinge on the transverse interfaces. A similar situation is illustrated in Figure 6 for grains containing transverse marker lines. In this case, atoms are deposited in diffusion creep at the transverse boundary between the two grains on the right, and this leads again to marker offsets in diffusion creep but no offsets in Harper–Dorn creep. Measurements of the offsets in marker lines, therefore, provide a unique opportunity for distinguishing between these two creep mechanisms.^[31]

Although grain-boundary sliding occurs as an accommodation process in diffusion creep, and the appearance of the boundary offsets is identical to those occurring in conventional Ratchinger sliding, it is important to note there is a true physical distinction between Ratchinger sliding and the

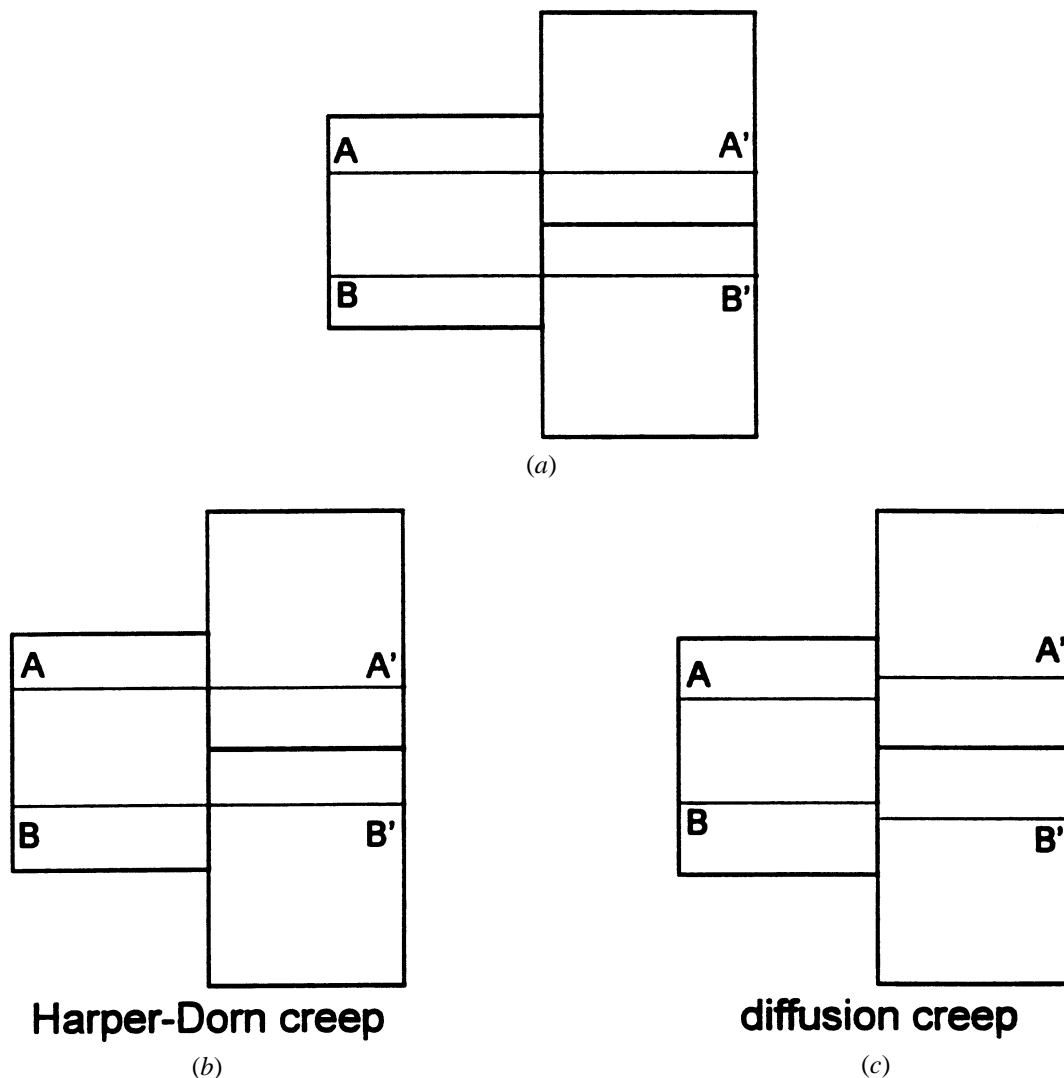


Fig. 6—Appearance of three grains in a polycrystalline matrix (a) before tensile creep with transverse marker lines AA' and BB', (b) after Harper–Dorn creep, and (c) after diffusion creep: the tensile axis is vertical.

sliding serving to accommodate diffusion creep. Thus, in Rachinger sliding the grains remain essentially equiaxed, and they become displaced with respect to each other so that there is a net increase in their number lying along the tensile axis; whereas in diffusion creep the individual grains become elongated through diffusive flow, the sliding serves only to accommodate this change in shape, the grains retain their relative positions within the polycrystalline matrix, and there is no net increase in the number of grains lying along the tensile axis. In order to clearly distinguish between these two types of sliding,^[17] the accommodating sliding in diffusion creep is termed Lifshitz grain-boundary sliding.^[33]

V. UNAMBIGUOUS EVIDENCE FOR THE OCCURRENCE OF DIFFUSION CREEP AND HARPER–DORN CREEP

In view of the many claims and counterclaims regarding the occurrence or absence of diffusion creep and Harper–Dorn creep, it is instructive to examine whether there are any experiments providing unique and unambiguous evidence for the occurrence of either mechanism. As will be

demonstrated in this section, there is good evidence supporting both of these creep mechanisms as viable deformation processes during creep at very low stress levels.

A. Diffusion creep

Equation [1] defines Nabarro–Herring diffusion creep when $n = 1$, $p = 2$, and $D = D_\ell$, but there is some uncertainty regarding the precise value of the dimensionless constant, A . In the theoretical derivation of the diffusion creep model, Herring^[11] obtained $A = 13.3$ for polycrystals tested in uniaxial tension and having fully-relaxed grain boundaries, but subsequently, it was shown that the value of A may lie in the range of 12 to 40 depending on the grain shape and testing conditions.^[34] A very detailed examination of creep data has suggested that the most appropriate value is $A \approx 40$ under tensile creep conditions,^[35] thereby demonstrating that the experiments tend to give creep rates that are a little faster, by up to a factor of two, than those expected from the theoretical analysis. Nevertheless, the consistency suggests there is generally reasonably good agreement with the diffusion creep model.



Fig. 7—Appearance of a Mg-0.55 pct Zr alloy after diffusion creep to a strain of 13.3 pct at 673 K: the tensile axis is vertical, denuded zones are visible along many of the transverse boundaries, and there are offsets in the hydride stringers at points where they impinge on the transverse grain boundaries.^[32]

Only three sets of measurements have been taken to confirm the predictions of Figures 5 and 6 regarding the offsets incurred through Lifshitz sliding in diffusion creep. In the first report, a Mg-0.55 pct Zr alloy with a grain size of $\sim 80 \mu\text{m}$ was tested to a strain of 13.3 pct at 673 K under an applied stress of 2 MPa, and measurements were taken to determine the contribution of sliding to the total strain, ξ , using the offsets introduced at transverse grain boundaries in hydride stringers lying approximately parallel to the tensile axis.^[36] This sample is illustrated in Figure 7 where the tensile axis is vertical, and there are well-defined denuded zones lying preferentially along the transverse grain boundaries.^[32] The results from these measurements gave $\xi \approx 60$ pct, and subsequently there were reports of $\xi \approx 50$ pct in a Mg-0.62 pct Mn alloy^[37] and $\xi \approx 51$ pct in a Mg-0.55 pct Zr alloy^[38] where both of these other materials were tested under conditions appropriate to diffusion creep. It is important to note these three measurements are mutually consistent, and they give a value of $\xi \approx 50$ to 60 pct in diffusion creep. In practice, however, Lifshitz sliding and diffusion creep are complementary processes that cannot be considered to make separate contributions to the total creep strain.^[39] This means that attempts to divide these contributions^[40,41] are meaningless because they depend uniquely upon the precise definitions of strain incorporated into the analysis.^[39] It has been shown that measurements of ξ based

on offsets in longitudinal markers have an upper-bound limiting value of $\xi \approx 50$ to 70 pct so that the measurements of ξ in diffusion creep are consistent with the expectations for diffusion creep as documented in Figure 5.

The mechanical characteristics of the sample depicted in Figure 7 were analyzed in detail by Harris,^[35] and it was shown that all of the data for this sample are consistent with Nabarro–Herring diffusion creep occurring as the dominant flow process. The result of $\xi \approx 60$ pct is also consistent with diffusion creep but, nevertheless, it has been suggested this experimental value of ξ may result from the occurrence of Rachinger sliding rather than Lifshitz sliding.^[30,42] This latter suggestion can be checked by noting that Lifshitz sliding is associated with grain elongation whereas Rachinger sliding entails no elongation of the individual grains. Careful measurements showed that the average grain aspect ratio was ~ 1.25 in this sample,^[32] and this is consistent with the experimental total strain of 13.3 pct which requires a grain aspect ratio of ~ 1.28 . It is also consistent with measurements of the denuded zones^[35] which suggest a grain aspect ratio of ~ 1.19 . All of these measurements are, therefore, mutually consistent, and they provide very strong evidence for the occurrence of Nabarro–Herring diffusion creep in this material. Conversely, the results cannot be interpreted either in terms of Harper–Dorn creep where there are no anticipated offsets in marker lines or Rachinger sliding where there is no grain elongation.

B. Harper–Dorn Creep

Many experimental results have been attributed to Harper–Dorn creep, but a careful analysis shows this attribution is often in error.^[43,44] Nevertheless, an examination of the original data of Harper and Dorn,^[13] when supplemented by the later experimental data of Harper *et al.*,^[45] provides strong evidence supporting the occurrence of some unique creep mechanism at these very low stress levels.

Figure 8 shows the original data of Harper and Dorn^[13] obtained at 920 K and plotted logarithmically as the steady-state creep rate against the applied stress for polycrystalline pure Al with a grain size of 3 mm and for a single crystal. At the higher stresses, there is a value of $n = 4.5$ which is consistent with dislocation creep in aluminum. At the lower stresses, the experimental points appear to lie along a line having a slope of $n = 1$, where the position of this line is more than two orders of magnitude faster than anticipated for Nabarro–Herring diffusion creep with $d = 3$ mm. As noted earlier, the consistency between experimental creep rates in metals and the predictions of Nabarro–Herring creep is generally to within a factor of ~ 2 . Furthermore, by linearly extrapolating to lower stresses the line for $n = 4.5$ and dislocation creep, it is apparent from Figure 8 that the experimental datum points recorded at the two lowest stresses cannot relate to dislocation creep because Nabarro–Herring diffusion creep is the faster process at these stress levels. This discrepancy between the experimental points and the predictions of diffusion creep is surprising, especially for pure aluminum where the diffusion coefficient is well established, and it provides strong support for the advent of a new and different creep mechanism under these experimental conditions.

Further evidence for a new mechanism is provided by the later experiments of Harper *et al.*,^[45] also conducted on the

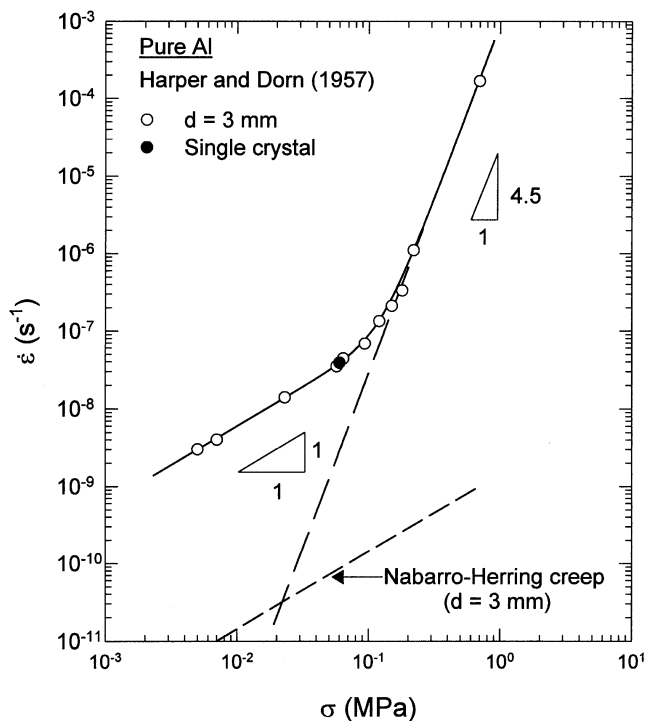


Fig. 8—Steady-state creep rate vs stress from the results of Harper and Dorn;^[13] also included are the prediction for Nabarro-Herring creep for a grain size of $d = 3$ mm and a linear extrapolation of the dislocation creep line where $n = 4.5$.

same pure aluminum at 920 K, where transverse marker lines were scribed on the samples prior to testing, and measurements were taken of the boundary offsets after creep in order to determine the contribution from grain-boundary sliding, ξ . The results show a peak value of ξ in the vicinity of the transition from $n \approx 4.5$ to $n = 1$ but very low, and decreasing, values of ξ within the Harper-Dorn region. Thus, the values of ξ decreased from 14.9 pct at $\sigma = 9.3 \times 10^{-2}$ MPa to 5.1 pct at $\sigma = 2 \times 10^{-2}$ MPa. These very low values of ξ are not consistent with the occurrence of Lifshitz sliding in diffusion creep, as illustrated schematically in Figure 6(c). On the contrary, the results are consistent with Figure 6(b), and they support the proposal that Harper-Dorn creep is a new and different creep mechanism.

Although the evidence in Figure 8, when combined with the later measurements of ξ , provides strong support for a different creep mechanisms at very low stresses, nevertheless, the situation is more complex because there is good experimental evidence showing that Harper-Dorn creep is observed in pure Al only when the initial dislocation density is low ($\sim 10^7$ – 3×10^8 m $^{-2}$).^[46] Furthermore, this is consistent with the experimental observation that the dislocation density remains constant, independent of the applied stress, and equal to a value of $\sim 5 \times 10^7$ m $^{-2}$ within the Harper-Dorn creep region.^[47] It is also significant to note that a dislocation density of this magnitude generates random internal stresses of the order of $\sim 2 \times 10^{-2}$ MPa,^[29] which is a little lower than the stress marking the transition to Harper-Dorn creep, as documented in Figure 8.

It was suggested very recently that the relatively rapid strain rates associated with Harper-Dorn creep may be a direct consequence of increases in creep rate due to the

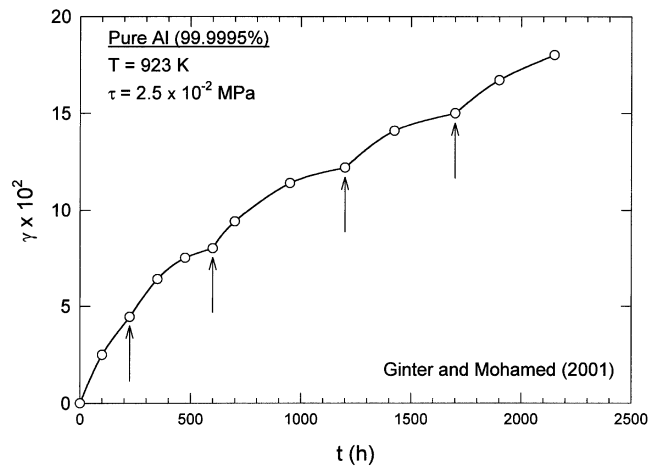


Fig. 9—Possible evidence for the occurrence of dynamic recrystallization during the creep of very high purity (99.9995 pct) aluminum when tested at 923 K under a shear stress of 2.5×10^{-2} MPa within the region attributed to Harper-Dorn creep.^[49]

occurrence of dynamic recrystallization during creep.^[48,49] An experimental example of this effect is shown in Figure 9 for very high purity (99.9995 pct) Al tested at 923 K under a shear stress, τ , of 2.5×10^{-2} MPa, where the plot shows the shear strain, γ , as a function of the testing time, t .^[49] If this effect occurred in the early experiments of Harper and Dorn,^[13] it is possible the measured creep rates may be erroneously high due to these perturbations in the creep curves. For example, by determining the average of the minimums occurring in the creep rates following jumps of the type shown in Figure 9, it was demonstrated that these datum points fall along a line of slope ~ 2.5 instead of ~ 1 , as shown in Figure 10 where the shear strain rate, $\dot{\gamma}$, is plotted against the shear stress, τ .^[48] It is important to note, however, that these minimum rates are also faster than those anticipated by extrapolation for dislocation creep, thereby again suggesting that a different creep mechanism occurs at these lowest stress levels.

In practice, close inspection suggests several reasons for rejecting any explanation of Harper-Dorn creep based on dynamic recrystallization. First, the jumps in the creep curve shown in Figure 9 are not clearly defined, and indeed the same experimental datum points were reported in an earlier publication where the results were plotted as a smooth and continuous curve (Figure 3(b) of Reference 46). Second, dynamic recrystallization occurs during the creep of very high-purity materials at regular and very well-defined strain increments,^[50,51] whereas the incremental strains recorded between each jump in Figure 9, and especially in other similar plots,^[48] tend to be relatively nonuniform. Third, the original experiments of Harper and Dorn,^[13] including also those of Harper *et al.*,^[45] were conducted using aluminum of 99.99 pct purity, and there is no evidence for the occurrence of dynamic recrystallization in aluminum of this purity.^[48,52] Fourth, there is also no experimental evidence for the occurrence of dynamic recrystallization in very high-purity Al at the very high temperatures and low stresses associated with experiments of the type documented in Figure 9; in practice, grain growth may be a more appropriate restoration mechanism under these conditions.^[53,54] Fifth, any attempt to attribute the relatively rapid strain rates

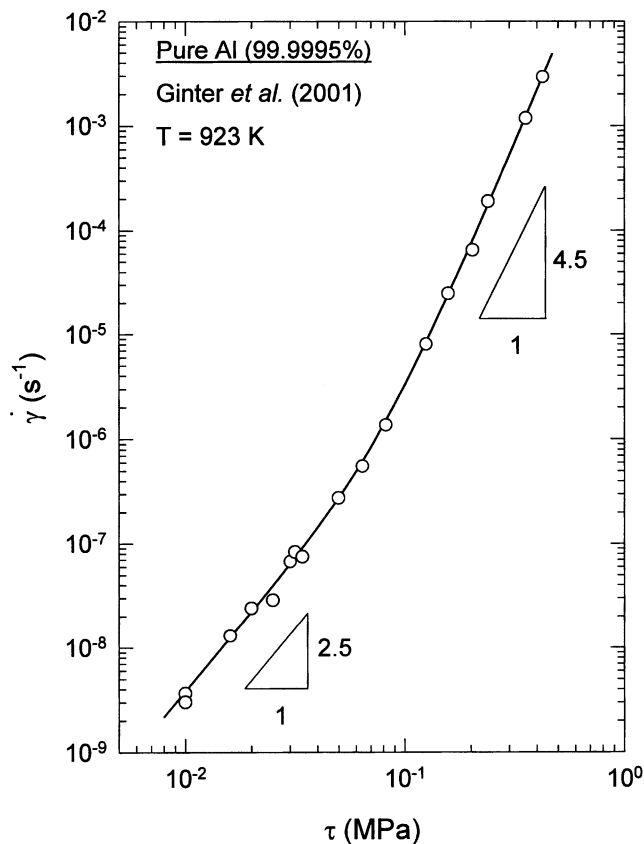


Fig. 10—Shear strain rate vs shear stress for very high purity (99.9995 pct) aluminum showing the change in slope to a value of $n \approx 2.5$ at low stresses when taking the minimum creep rates following apparent bursts of dynamic recrystallization.^[48]

recorded in the earlier experiments^[13,45] to either a failure to reach a steady-state condition or the advent of dynamic recrystallization during creep must also necessarily account for the very low, and decreasing, values of ξ recorded within the region of Harper–Dorn creep which clearly preclude the occurrence of diffusion creep.

In summary, the arguments against the early results of Harper and Dorn^[13] are not well-founded and there remains no explanation for the low values of ξ reported in the detailed experiments of Harper *et al.*^[45] Furthermore, no attempts have been made to measure ξ in any subsequent investigations of Harper–Dorn creep despite the fact that the values of ξ provide an important parameter in distinguishing between Harper–Dorn creep and diffusion creep.

VI. DISCUSSION

The evidence for Nabarro–Herring diffusion creep is now strong. In particular, the very detailed and extensive measurements performed on a single sample of a Mg-0.55 pct Zr alloy provide very good support for the occurrence of this mechanism. These measurements include the overall creep rate,^[35] the width of the denuded zones,^[35] an estimate of the contribution of Lifshitz sliding from offsets measurements,^[36] and a determination of the average grain aspect ratio.^[32] All these measurements are mutually consistent, and, when taken together, they do not support the occurrence of any other known creep mechanism. In addition, the recent

use of advanced techniques, such as atomic force microscopy, has provided clear and essentially unambiguous evidence for the occurrence of diffusion creep in Cu^[55] and in a Mg-0.5 pct Zr alloy.^[56]

It is important to note that numerous arguments have been presented both supporting^[28,57–59] and opposing^[30,42,60] a relationship between the presence of denuded zones and the occurrence of diffusion creep. Clearly, the appearance of the denuded zones in Figure 7 is consistent with the advent of diffusion creep, as depicted in Figure 2(b). Nevertheless, there are several observations of denuded zones in samples subjected to an annealing treatment but without any creep testing,^[38,61,62] thereby demonstrating unequivocally that, as also concluded elsewhere,^[32,63] the presence of denuded zones cannot be taken as conclusive evidence for the occurrence of diffusion creep.

The evidence for Harper–Dorn creep is less definitive than for diffusion creep, but nevertheless the early results of Harper and Dorn,^[13] when combined with the later microstructural observations by Harper *et al.*,^[45] provide very strong support for the proposal that Harper–Dorn creep is a unique creep mechanism occurring under some limited testing conditions. In particular, the detailed measurements of the sliding contribution, ξ , demonstrate conclusively that the pure aluminum used in these experiments was not deforming at the very lowest stresses by conventional diffusion creep despite the fact that, as illustrated in Figure 8, Nabarro–Herring creep would be expected under these testing conditions through a simple extrapolation of the line for dislocation creep.

There have been some very recent attempts to replicate the early experiments of Harper and Dorn^[13] using samples of pure Al but these attempts have been unsuccessful.^[27,64] One significant difficulty in performing these experiments is that the creep rates associated with Harper–Dorn creep in samples of pure Al are extremely slow. This problem may be partially overcome by performing the experiments on Al–Mg alloys. The effect of adding Mg to aluminum is illustrated schematically in Figure 11, where it is apparent that the addition of Mg in solid solution displaces the line for dislocation creep to higher stresses and, since the line for Harper–Dorn creep remains essentially invariant, the maximum strain rate associated with Harper–Dorn creep is, therefore, effectively displaced to faster rates by up to one order of magnitude. Harper–Dorn creep has been documented in an Al-5 pct Mg alloy,^[47,65] and this would appear to be an ideal candidate material for future investigations of the occurrence of Harper–Dorn creep.

VII. SUMMARY AND CONCLUSIONS

1. Logarithmic plots of the steady-state creep rate vs the applied stress generally reveal an extensive region where the stress exponent is high, typically in the range of ~ 3 to 6, but with a transition at very low stresses to a region where the stress exponent is low and of the order of ~ 1 to 2. The behavior at low stresses is generally attributed to the occurrence of a new creep mechanism, such as grain-boundary sliding, diffusion creep, and/or Harper–Dorn creep.
2. Several recent reports have suggested that diffusion creep

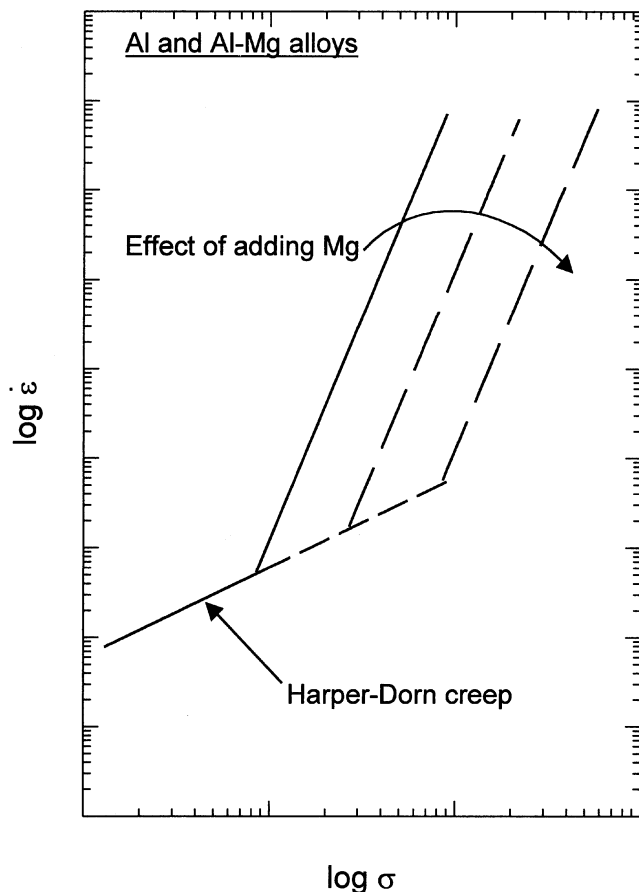


Fig. 11—A schematic illustration of the potential for increasing the maximum strain rate associated with Harper–Dorn creep by performing experiments on Al-Mg solid solution alloys.

and Harper–Dorn creep may not be viable creep mechanisms. The validity of this suggestion was evaluated by re-examining selected creep data in the low stress region.

3. It is concluded there is good experimental evidence for the occurrence of both diffusion creep and Harper–Dorn creep under at least some experimental conditions.

ACKNOWLEDGMENTS

This work was supported by the United States Department of Energy under Grant No. DE-FG03-92ER45472.

REFERENCES

1. P. Phillips: *Proc. Phys. Soc. London*, 1903–1905, vol. 19, pp. 491–511.
2. E.N. daC. Andrade: *Proc. R. Soc. A*, 1910, vol. 84, pp. 1–12.
3. E.N. daC. Andrade: *Proc. R. Soc. A*, 1914, vol. 90, pp. 329–42.
4. J.E. Bird, A.K. Mukherjee, and J.E. Dorn: in *Quantitative Relation Between Properties and Microstructure*, D.G. Brandon and A. Rosen, eds., Israel Universities Press, Jerusalem, 1969, pp. 255–342.
5. F.A. Mohamed and T.G. Langdon: *Acta Metall.*, 1974, vol. 22, pp. 779–88.
6. P. Yavari and T.G. Langdon: *Acta Metall.*, 1982, vol. 30, pp. 2181–96.
7. W.D. Nix and B. Ilshner: in *Strength of Metals and Alloys (ICSMA 5)*, P. Haasen, V. Gerold, and G. Kostorz, eds., Pergamon, Oxford, United Kingdom, 1980, vol. 3, pp. 1503–30.
8. O.D. Sherby and P.M. Burke: *Prog. Mater. Sci.*, 1967, vol. 13, pp. 325–90.
9. S.V. Raj and T.G. Langdon: *Acta Metall.*, 1989, vol. 37, pp. 843–52.

10. F.R.N. Nabarro: *Report of a Conference on Strength of Solids*, The Physical Society, London, 1948, pp. 75–90.
11. C. Herring: *J. Appl. Phys.*, 1950, vol. 21, pp. 437–45.
12. R.L. Coble: *J. Appl. Phys.*, 1963, vol. 34, pp. 1679–82.
13. J. Harper and J.E. Dorn: *Acta Metall.*, 1957, vol. 5, pp. 654–65.
14. F.A. Mohamed and J. Wolfenstine: in *Hot Deformation of Aluminum Alloys*, T.G. Langdon, H.D. Merchant, J.G. Morris, and M.A. Zaidi, eds., TMS, Warrendale, PA, 1991, pp. 223–37.
15. T.G. Langdon and R.B. Vastava: in *Mechanical Testing for Deformation Model Development*, R.W. Rhode and J.C. Swearingen, eds., ASTM, Philadelphia, PA, 1982, Special Technical Publication STP765, pp. 435–51.
16. T.G. Langdon: *Mater. Sci. Eng.*, 1994, vol. A174, pp. 225–30.
17. W.R. Cannon: *Phil. Mag.*, 1972, vol. 25, pp. 1489–97.
18. W.A. Rachinger: *J. Inst. Met.*, 1952–53, vol. 81, pp. 33–41.
19. T.G. Langdon: *Acta Metall. Mater.*, 1994, vol. 42, pp. 2437–43.
20. F.A. Mohamed and T.G. Langdon: *Scripta Metall.*, 1976, vol. 10, pp. 759–762.
21. P.K. Chaudhury and F.A. Mohamed: *Acta Metall.*, 1988, vol. 36, pp. 1099–1110.
22. P.K. Chaudhury, V. Sivaramakrishnan and F.A. Mohamed: *Metall. Trans. A*, 1988, vol. 19A, pp. 2741–2752.
23. R.Z. Valiev and T.G. Langdon: *Acta Metall.*, 1993, vol. 41, pp. 949–54.
24. L.K.L. Falk, P.R. Howell, G.L. Dunlop and T.G. Langdon: *Acta Metall.*, 1986, vol. 34, pp. 1203–1214.
25. B. Wilshire: *Proc. 4th Int. Conf. on Creep and Fracture of Engineering Materials and Structures*, B. Wilshire and R.W. Evans, eds., The Institute of Metals, London, 1990, pp. 1–9.
26. O.A. Ruano, J. Wadsworth, J. Wolfenstine, and O.D. Sherby: *Mater. Sci. Eng.*, 1993, vol. A165, pp. 133–41.
27. W. Blum and W. Maier: *Phys. Status Solidi (a)*, 1999, vol. 171, pp. 467–74.
28. B. Burton and G.L. Reynolds: *Mater. Sci. Eng.*, 1995, vol. A191, pp. 135–41.
29. F.R.N. Nabarro: *Phys. Status Solidi (a)*, 2000, vol. 182, pp. 627–29.
30. J. Wadsworth, O.A. Ruano, and O.D. Sherby: in *Creep Behavior of Advanced Materials for the 21st Century*, R.S. Mishra, A.K. Mukherjee, and K.L. Murty, eds., TMS, Warrendale, PA, 1999, pp. 425–39.
31. T.G. Langdon: *Scripta Mater.*, 1996, vol. 35, pp. 733–37.
32. T.G. Langdon: *Mater. Sci. Eng.*, 2000, vol. A283, pp. 266–73.
33. I.M. Lifshitz: *Sov. Phys. JETP*, 1963, vol. 17, pp. 909–20.
34. G.B. Gibbs: *Mém. Sci. Rev. Mét.*, 1965, vol. 62, pp. 781–86.
35. J.E. Harris: *Met. Sci. J.*, 1973, vol. 7, pp. 1–6.
36. R.C. Gifkins and T.G. Langdon: *Scripta Metall.*, 1970, vol. 4, pp. 563–67.
37. I.G. Crossland and J.C. Wood: *Phil. Mag.*, 1975, vol. 31, pp. 1415–19.
38. E.H. Aigeltinger and R.C. Gifkins: *J. Mater. Sci.*, 1975, vol. 10, pp. 1889–1903.
39. R.C. Gifkins, T.G. Langdon, and D. McLean: *Met. Sci.*, 1975, vol. 9, pp. 141–44.
40. B.-N. Kim and K. Hiraga: *Scripta Mater.*, 2000, vol. 42, pp. 451–56.
41. S.S. Sahay and G.S. Murty: *Scripta Mater.*, 2001, vol. 44, pp. 841–45.
42. O.A. Ruano, O.D. Sherby, J. Wadsworth, and J. Wolfenstine: *Scripta Mater.*, 1998, vol. 38, pp. 1307–14.
43. M.Z. Berbon and T.G. Langdon: *J. Mater. Sci. Lett.*, 1996, vol. 15, pp. 1664–66.
44. M.Z. Berbon, D.M. Owen, and T.G. Langdon: *The Johannes Weertman Symp.*, R.J. Arsenault, D. Cole, T. Gross, G. Kostorz, P.K. Liaw, S. Parameswaran, and H. Sizek, eds., TMS, Warrendale, PA, 1996, pp. 145–54.
45. J.G. Harper, L.A. Shepard, and J.E. Dorn: *Acta Metall.*, 1958, vol. 6, pp. 509–18.
46. F.A. Mohamed and T.J. Ginter: *Acta Metall.*, 1982, vol. 30, pp. 1869–81.
47. P. Yavari, D.A. Miller, and T.G. Langdon: *Acta Metall.*, 1982, vol. 30, pp. 871–79.
48. T.J. Ginter, P.K. Chaudhury, and F.A. Mohamed: *Acta Mater.*, 2001, vol. 49, pp. 263–72.
49. T.J. Ginter and F.A. Mohamed: *Mater. Sci. Eng.*, 2001, in press.
50. R.C. Gifkins: *J. Inst. Met.*, 1958–59, vol. 87, pp. 255–61.
51. G.J. Richardson, C.M. Sellars, and W.J.McG. Tegart: *Acta Metall.*, 1966, vol. 14, pp. 1225–36.
52. H. Yamagata: *Acta Metall. Mater.*, 1995, vol. 43, pp. 723–29.
53. H.J. McQueen, W. Blum, S. Straub, and M.E. Kassner: *Scripta Metall. Mater.*, 1993, vol. 28, pp. 1299–1303.

54. M.E. Kassner, J. Pollard, E. Evangelista, and E. Cerri: *Acta Metall. Mater.*, 1994, vol. 42, pp. 3223-30.
55. K.R. McNee, G.W. Greenwood, and H. Jones: *Scripta Mater.*, 2001, vol. 44, pp. 351-57.
56. K.R. McNee, H. Jones, and G.W. Greenwood: *Proc. 9th Int. Conf. on Creep and Fracture of Engineering Materials and Structures*, J.D. Parker, ed., The Institute of Materials, London, 2001, pp. 3-13.
57. G.W. Greenwood: *Scripta Metall. Mater.*, 1994, vol. 30, pp. 1527-30.
58. L. Kloc: *Scripta Mater.*, 1996, vol. 35, pp. 539-41.
59. G.W. Greenwood: *Creep Behavior of Advanced Materials for the 21st Century*, R.S. Mishra, A.K. Mukherjee, and K.L. Murty, eds., TMS, Warrendale, PA, 1999, pp. 413-23.
60. J. Wolfenstine, O.A. Ruano, J. Wadsworth, and O.D. Sherby: *Scripta Metall. Mater.*, 1993, vol. 29, pp. 515-20.
61. B.W. Pickles: *J. Inst. Met.*, 1967, vol. 95, pp. 333-37.
62. W. Vickers and P. Greenfield: *J. Nucl. Mater.*, 1967, vol. 24, pp. 249-60.
63. J.B. Bilde-Sørensen and D.A. Smith: *Scripta Metall. Mater.*, 1994, vol. 30, pp. 383-86.
64. K.R. McNee, H. Jones, and G.W. Greenwood: *Proc. 9th Int. Conf. on Creep and Fracture of Engineering Materials and Structures*, J.D. Parker, ed., The Institute of Materials, London, 2001, pp. 185-95.
65. P. Yavari, F.A. Mohamed, and T.G. Langdon: *Acta Metall.*, 1981, vol. 29, pp. 1495-1507.